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ROYAL AIRCRAFT ESTABLISHMENT

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SOME OBSERVATIONS ON THE

DEFORMATION OF GRAINS AND BOUNDARIES IN PURE ALUMINIUM AND THREE BINARY ALUMINIUM ALLOYS

by

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Some observations on the deformation of grains and boundaries in pure aluminium and three binary aluminium alloys

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Mat M1/10187/PJEF and JCT

SUMMARY

A small straining device has been constructed which can apply a tensile stress to an electropolished specimen at room temperature, or if required, immersed in liquid nitrogen.

With this device variations in deformation behaviour with temperature have been investigated. The effect of grain size has also been studied and comparisons made with the deformation produced by fatigue stresses.

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1 Introduction

Recent work on the fatigue of light alloys[1] has shown that there is often a fine balance between the tendency for grain boundary and transcrystalline fracture. It has been particularly noticed that some commercial alloys having a mixed grain size fail by an intercrystalline path in regions of small grains and by a transcrystalline path across large grains when subjected to corrosion fatigue. To a certain extent this may be due to the fact that regions of small grain size may be associated with segregation of alloying elements, and the presence of a higher concentration of some of these elements may cause boundary weakness. On the other hand it was considered that from the known facts of deformation in metals one might expect different behaviour purely on a grain size basis. It is well known that non-homogeneous deformation such as kinking is favoured where free glide is restricted. Also that the freedom to glide is dependent on the mean free path of dislocations released by Frank-Read sources[2] which is controlled mainly by grain size. In fact any impedance to glide within the grains might be expected to transfer the deformation to the boundary regions. We should at this stage differentiate between boundary glide which is usually associated with creep conditions and is closely concentrated along the boundaries, and the heavy curvatures produced in the boundary regions under more normal rates of loading with which we are mainly concerned in this work.

To test the hypothesis that deformation behaviour depends on the freedom of movement of large numbers of dislocations, i.e. by the activaties of Frank-Read sources tests were made varying grain size, testing at room and liquid nitrogen temperature. The materials used were pure aluminium and some simple alloys. Comparisons were also made with specimens of the same materials which had been fatigued.

2 Apparatus

The apparatus consisted of a straining device which is illustrated in Fig. 1 which could be immersed in liquid nitrogen. The specimens were stamped from 0.025" thick strip previously cold rolled from the forged billet. These specimens were of the same form as those used for the fatigue investigation. They were in all cases strained to fracture, and due to their varying section different degrees of strain could be observed along their length.

3 Materials

The materials used were 'Super purity aluminium', aluminium 1% manganese, aluminium 10% zinc and aluminium 4% copper. All the alloys were made from high purity components the aluminium in each case being 'super pure'.

Heat Treatments

Aluminium - Large grain size (Average 1 mm) cold rolled and annealed 1 hour at 620°C.

Small grain size (Average 0.1 mm) cold rolled and annealed 15 minutes at 350°C.

Aluminium 1% Manganese - Solution heat treated at 620°C and cold water quench.

Aluminium 10% Zinc - Solution heat treated 1 hour at 420°C then cold water quench. (All tests were made immediately after quenching).

Aluminium 4% copper - Solution heat treat 1 hour at 520°C then cold water quench.

Specimen Preparation

All specimens were stamped from rolled strip, heat treated then electropolished from the 'as rolled' surface, no intermediate grinding being employed.

4 Microscopical examination

Room and low temperature (liquid nitrogen) tests
Pure Aluminium: Large grains (1 mm average)

The room temperature tensile pull on this material (as illustrated in Figs. 2 and 4) produced characteristic slip bands; in some grains on three sets of planes. The bands are generally straight and practically continuous across the grains from one boundary to another. The bottom left hand grain in Fig. 2 has impeded slip in the right hand grain in the neighbourhood of the mutual grain boundary. Figs. 3 and 5 show similar specimens tested at liquid nitrogen temperature. A great deal of cross slip is evident due to the fact that free glide is impeded at low temperatures i.e. the activation of Frank-Read sources is more difficult. Fig. 5 shows a grain in which slip bands are less continuous due presumably to the greater effectiveness of barriers in producing piling-up of dislocations at low temperatures. The grain in the bottom left hand corner has been deformed by a characteristic rumpled pattern often found near grain boundaries where free glide has been obstructed.

Small grain size (0.1 mm average)

Fig. 6 shows a small grain size aluminium specimen which has been strained. It will be seen that surface furrowing and rumpling has occurred particularly near the grain boundaries. The slip bands themselves are not straight presumably because of the occurrence of cross slip. Fig. 7 shows a similar grain size specimen strained at liquid nitrogen temperature. By comparison with Fig. 6 it will be seen that the slip is more irregular, and even more marked boundary deformation has occurred. The cellular network pattern observed in Fig. 5 is also present. Fig. 8 shows another region of the room temperature tested specimen where large and small grains occur together. It can be seen that the large grains have formed fairly long, straight slip bands whereas the top included small grain shows a complex cross slip pattern. A similar specimen tested at liquid nitrogen temperature (Fig. 9) shows very irregular slip including the cellular pattern, and far more deformation near grain boundaries.

Aluminium 1% Manganese alloy (grain size approximately 0.5 mm)

The aluminium 1% manganese alloy (Fig. 10) shows rather more irregular slip within the grains than pure aluminium and a certain amount of grain boundary disturbance. The liquid nitrogen temperature (-195°C) test shows a markedly different appearance from both the room temperature tested aluminium - 1% manganese and also the low temperature tested pure aluminium as shown in Figs. 3, 5 and 9. Coarse slip has been almost completely inhibited in this alloy when tested at liquid nitrogen temperature presumably by the combined effect of the solute atoms and the low temperature.

Aluminium 10% Zinc (grain size approximately 0.5 mm)

Due to the rapid rate of age hardening of this alloy tests were conducted as soon as possible after quenching from the solution heat treated condition. Even so it is certain that some ageing must have occurred before the test. Fig. 12 shows a typical structure after deformation. It will be seen that glide is less interrupted than in aluminium 1% manganese but very sharply defined grain boundary deformation has occurred. This deformation often produces small deformation bands at grain corners as shown by arrow A. The low temperature deformation is shown in Fig. 13. This sharply defined boundary glide which occurred at room temperature is shown at a higher magnification in Fig. 13 where a grain boundary, presumably obliquely oriented to the surface of the crystal has slipped. A similar boundary region is shown in Fig. 15, this material was deformed at liquid nitrogen temperature. It can be seen that the boundary deformation is now more widespread and involves clearly defined fine slip in each grain.

Aluminium 4% copper alloy

Figs. 16 and 17 show an aluminium 4% copper microstructure strained at room temperature and liquid nitrogen temperature respectively. Apart from the fact that slip bands are straighter and more continuous at room temperature than at low temperature, there is not the striking difference in behaviour noticed in the aluminium 10% zinc alloy.

Room and liquid nitrogen temperature tests

Cold rolled materials:

(1) Pure Aluminium: The general disorientation caused by cold rolling pure aluminium is evident by the irregular nature of the subsequent tensile deformation (See Fig. 18). Even so a number of fairly long slip bands have formed. This can be compared with the same material tested at liquid nitrogen temperature, which is illustrated in Fig. 19, where very few marked slip bands are formed.

(2) Aluminium 1% Manganese alloy

This material behaved similarly to pure aluminium slip being very irregular at both room temperature (Fig. 20) and low temperature (Fig. 21).

(3) Aluminium 10% zinc alloy

In the cold rolled condition this alloy shows very little difference in behaviour when strained at room or liquid nitrogen temperature. Figs. 22 and 23 show the two conditions.

(4) Aluminium 4% copper alloy

When strained at room temperature this alloy showed a number of irregular cross slip bands apart from the background deformation. When strained at liquid nitrogen temperature this was not evident, the only deformation observed was of an irregular granular nature. Fig. 24 and 25 show these conditions.

5 Comparisons with Fatigue deformation produced in the same materials

Fig. 26 shows large grained pure aluminium fatigued at room temperature and Fig. 27 shows small grain size aluminium fatigued at room temperature. This is comparable with Fig. 6. It will be seen that the fatigue stresses have not produced slip striations as shown in Fig. 26 but rather a number of regions of fine cellular network deformation which seems to be the fatigue equivalent of cross slip under static stresses. Fig. 27 also shows very localised deformation on each side of the boundary in the form of two furrows. This effect had been noticed in larger grain sized aluminium fatigued at higher frequencies. Fig. 29 shows included grains and is the fatigue equivalent of Fig. 8. Fig. 30 shows localised boundary deformation produced in aluminium - 1% manganese alloy by fatiguing at liquid nitrogen temperature. This is comparable with Fig. 11 which, was statically strained at the same temperature. No direct comparisons can be made with the fatigue or static strain behaviour of the other alloys as ageing effects become an important factor in their fatigue behaviour.

Discussion

When a shear stress is applied to a crystal it may cause movement of mobile dislocations in the lattice, and, if the stress is large enough it may also activate 'Frank'-Read type dislocation sources. These sources suddenly generate large numbers of dislocations which move along a few atomic planes. These large numbers of dislocations produce slip bands on the crystal surface which can be easily seen with the light microscope, but the smaller movements of the pre-existing mobile dislocations may not be clearly resolved except with the electron microscope. However, their presence is often revealed to the light microscope by the surface undulations which large numbers of these fine slip bands may cause[3].

The presence of grain boundaries and the interaction between neighbouring grains greatly modifies the appearance of the surface deformation. These obstructions to free glide produce kink bands in the grains[4,5] and often particularly heavy curvature of the crystal near the grain boundaries. The grain size of the material is therefore an important factor in the type of deformation present. In the extreme case of a single crystal, unrestricted glide may result in almost perfect straight slip bands. Similar effects can be obtained in large grained aggregates, the slip bands extending to or very near the grain boundaries. Any obstruction to the movement of dislocations will cause a pile-up which may result in the slip band stopping[6]. This may occur within the grain if an effective barrier is present. Obstructions to the free movement of dislocations may also result in cross slip as described by Cahn[7] as well as kinking and associated flexural glide. Boas[8] has shown that the deformation of aggregates may also result in glide on other than the normal planes. These extra degrees of freedom allow easier accommodation of the strain in boundary regions. Thus deformation of even recrystallised grain aggregates can be considered as extremely complex and inhomogeneous.

The heavy deformation of pure aluminium, as for example by cold rolling, results in considerable fragmentation of the original grains which may then be considered as a large number of small disoriented crystallites of relatively perfect lattice surrounded by walls of less perfect lattice[9] where most of the dislocations will be concentrated. Thus under heavy deformation the surface of the grains may reveal the presence of this structure. Naturally the production of such a structure will modify subsequent deformation. These modifications will be due to (i) the disorientation and (ii) the thick walls of concentrated dislocations both of which introduce effective barriers to slip of the Frank-Read type[10].

At advanced stages of deformation the original grain boundaries may lose their identity due to the disorientation of the grains, but in the early stages the balance between grain and boundary region deformation will to a certain extent depend on the freedom of movement of dislocations within the grain. If easy glide can occur in the grains then the grain boundary regions remain relatively undeformed. Small aggregates by virtue of the fact that they reduce the mean free glide path, and thus inhibit easy glide, will suffer from heavier boundary deformation. This heavier deformation in the boundary regions represents higher stresses transferred to the boundaries themselves.

At liquid nitrogen temperature these effects are very much more marked as the Frank-Read sources are less readily activated. Solute atoms behave in a similar manner by also inhibiting glide in the grains.

Under these conditions it has been shown that deformation in the boundary regions becomes more marked. It is of interest to note that although more boundary region deformation occurred in the aluminium 10% zinc alloy at low temperature, the flow of what is believed to be the depleted solid solution at the grain boundary itself, was not as marked.

Conclusions

1 Large grain size materials, particularly pure aluminium, show straighter and more continuous slip bands than smaller grain size materials.

2 Small grain size aluminium deforms more heavily in the grain boundary regions than a large grain size structure.

3 Lowering the deformation temperature accentuates these differences.

4 The above observations are true both under static and fatigue stresses.

5 Cross slip is more prevalent in specimens strained at low temperatures and may be further encouraged by the presence of solute atoms.

6 In the specific case of a material such as aluminium - 10% zinc which may contain grain boundaries depleted of solute atoms, grain boundary flow is more marked at room temperature than at liquid nitrogen temperature. This is in agreement with earlier work on this alloy which failed by intercrystalline fatigue fracture at room temperature and by a transcrystalline fracture at liquid nitrogen temperature.

7 Static deformation of cold rolled materials occurs mainly in the form of very short, irregular slip bands giving the specimen surface a roughened appearance. Again, low temperature inhibits the formation of long bands.

8 Aluminium 10% zinc alloy in the cold rolled condition shows a marked similarity of behaviour at room and liquid nitrogen temperature.

9 There is a marked similarity of behaviour under specific test conditions of fatigued and statically strained specimens.

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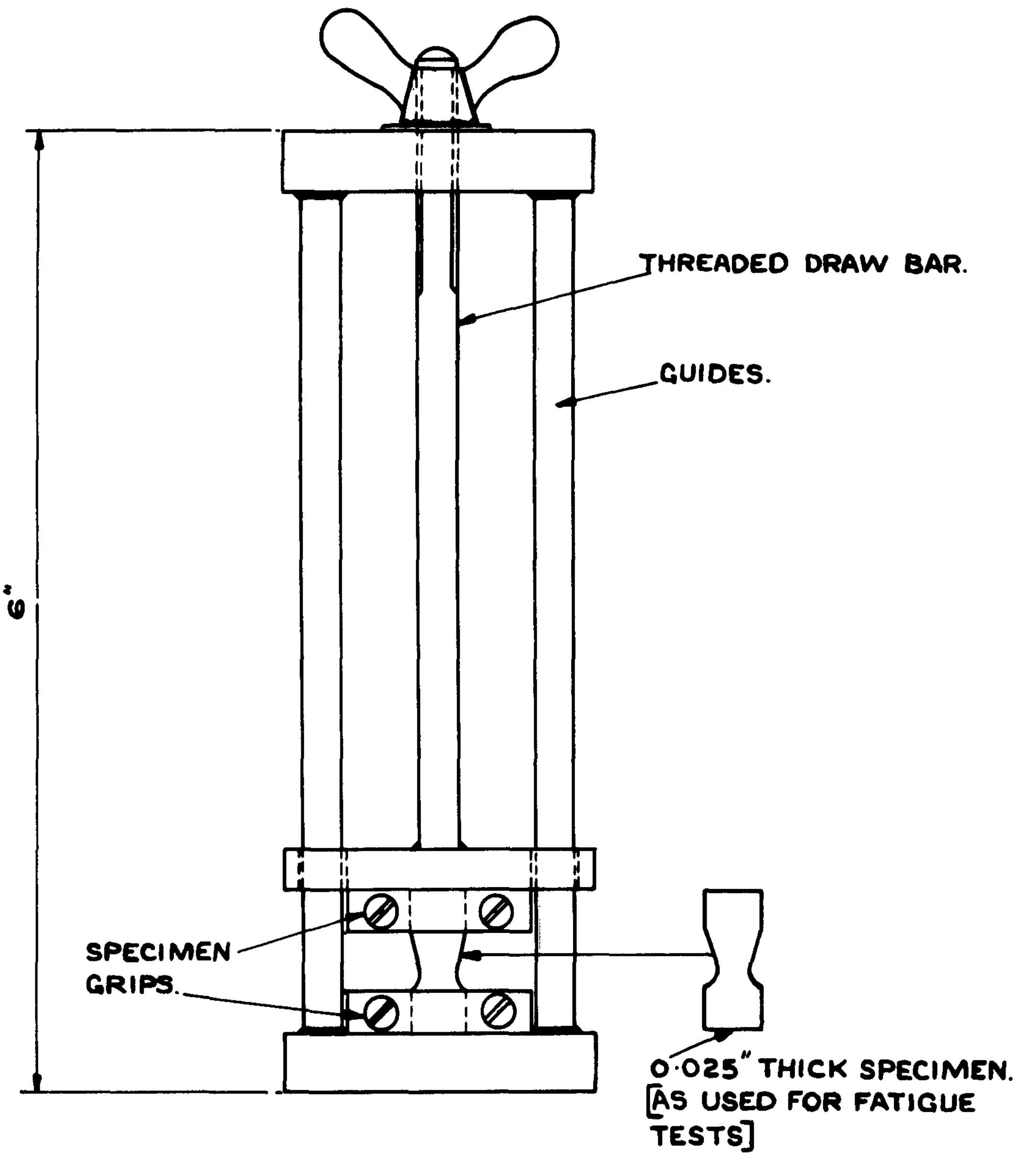
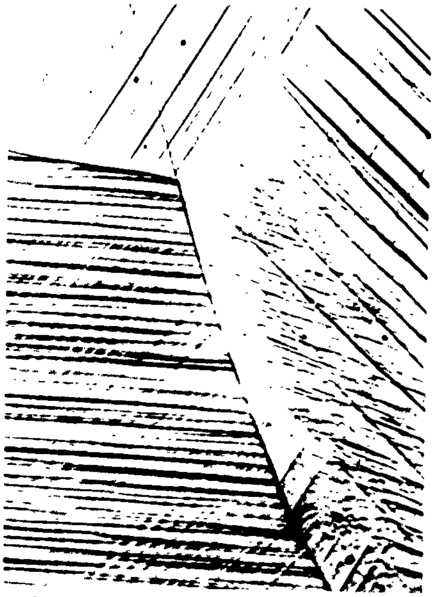


FIG. 1. STRAINING DEVICE FOR LOW TEMPERATURE TESTS.



2 ×250

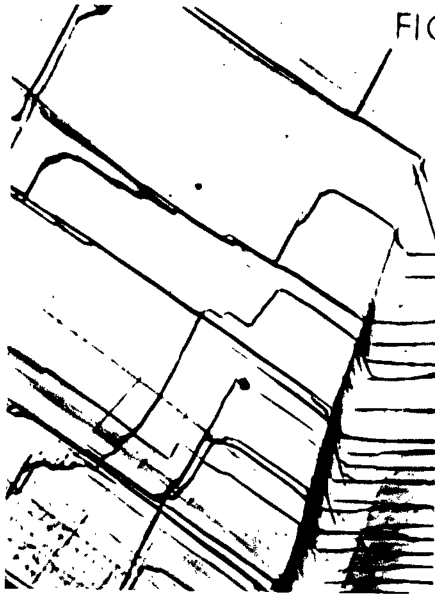
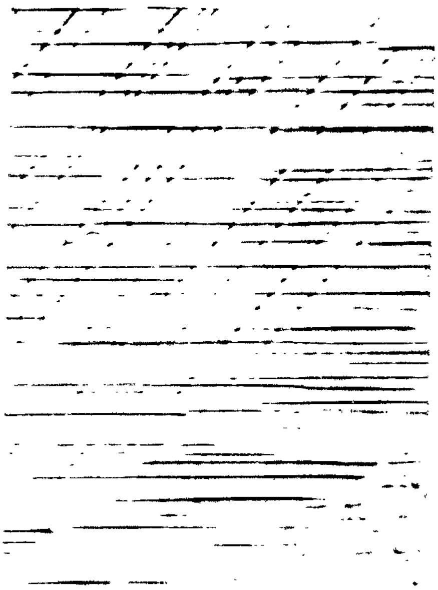


FIG 3 ×250



4 ×250

FIG 5 ×25

·250

·25

ROOM TEMPERATURE

LIQUID NITROGEN TEMPERATURE

FIG 2 - 7

FIG 8 x250

FIG 9 x250

FIG 10 x250

FIG 11 x250

FIG 12 x250

FIG 13 x250

ROOM TEMPERATURE

LIQUID NITROGEN TEMPERATURE

FIG 8 - 13

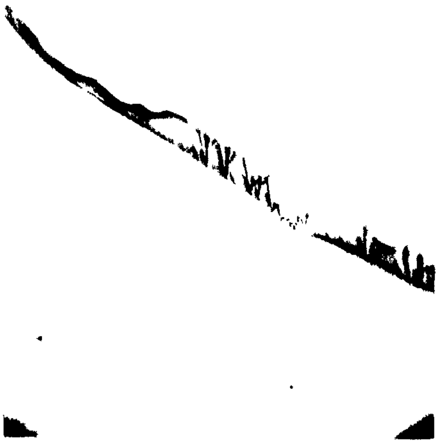


FIG 14 ×1000

FIG 15 ×1500

FIG 16 ×250

FIG 17 ×250

FIG 18 ×250

FIG 19 ×2[illegible]

ROOM TEMPERATURE

LIQUID NITROGEN TEMPERATURE

FIG 14 - 19

FIG 20 - 25

FIG 20 ×250

FIG 21 ×250

FIG 22 ×250

FIG 23 ×250

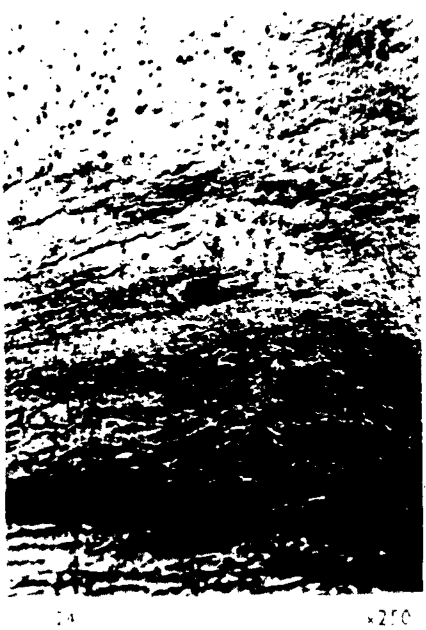


FIG 24 ×250

FIG 25 ×250

ROOM TEMPERATURE

LIQUID NITROGEN TEMPERATURE

FIG 20 - 25

FIG 26 x100

FIG 27 x250

FIG 28 x250

FIG 29 [illegible]

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